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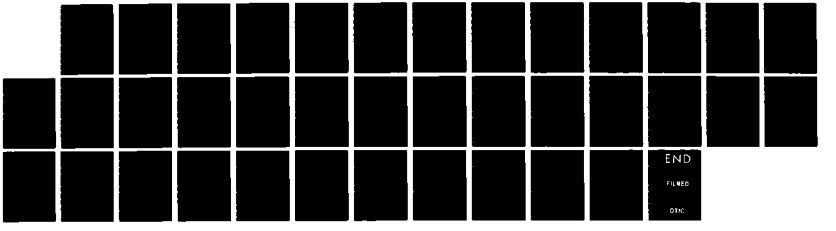

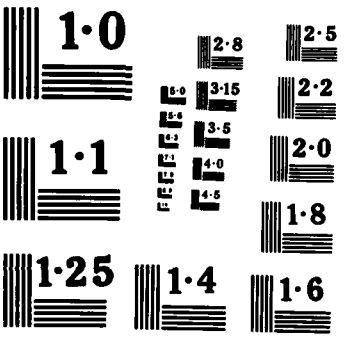

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ABSTRACT

Although there is mounting interest in the measurement of stresses in composite materials after fabrication and/or use, few measurements to date have taken into account the three dimensional nature of the stress system in such materials. Most data gives only the net stress, that is the difference between principal stresses. This report develops a procedure for a more complete measurement (in a reasonable time), including the separation of macrostresses and microstresses. If time does not permit a full investigation, measurements of the lattice parameters of the component phases provide a simple way to sample the hydrostatic component due to differential thermal contraction. The Barrett-Predecki method of adding filler is particularly promising for stress measurements in those composites whose component phases do not give appropriate diffraction peaks. This procedure could also be used for monitoring stresses during the useful life of such materials.

INTRODUCTION

There are a variety of origins for the residual stresses (and strains) in a composite. Firstly, there are the "macrostresses" that may arise due to fabrication - rolling, machining, etc. These occur because of the relative elongation of one region of the material relative to the other, usually the near surface regions relative to the bulk. There are also various sources of "microstresses": a) those stresses that arise due to differential cooling of the matrix and reinforcement, b) those due to their different plastic behavior, c) those due to compatibility strains because of their different elastic constraints, d) those due to coherency strains, e) those due to differences in constraints (surface vs. bulk). Especially if there is strong bonding between a fiber or particle and the matrix these microstresses may vary appreciably through the matrix, perhaps as shown in Fig. 1. In this review, the exact nature of these two types of stresses will be discussed first, and then how they effect the crystalline diffraction pattern. The use of x-ray diffraction has a long history in sampling stresses, with a particularly dynamic period in the last decade, so that its use is well established, and most details are understood in depth. More recently, neutron scattering has been shown to offer some advantages which will be examined here. Finally, the various techniques applicable to com-

posites will be reviewed.

It is perhaps worth mentioning, that the largest body of experience on the use of diffraction for sampling stress systems has been in the basic industries for measurements of stresses in steel products. Steel is, after all, a composite! Therefore, the procedures described here maybe of use in this field as well. Indeed, the origins of my own thinking on this topic lie with an interest in separately determining the stresses in the ferrite and carbide phases.

MACROSTRESSES ($^{m}\sigma_{ij}$):

The origin of macro-residual stresses may be envisioned as the process shown in Fig. 2. One piece of material is longer than another; this could be the near surface region, due to rolling, grinding, peening, or the action of one ply on another, etc. This piece is compressed by applying traction, it is joined to the bulk, and then the traction is removed. It attempts to spring back to its original dimensions, but is resisted by the bulk, and applies a tensile stress to this bulk. Thus it is in compression, and puts the bulk in tension, Fig. 3. The first fundamental point about these macrostresses is that the integration through the thickness must be zero for equilibrium (Mura,

1982; Noyan, 1983). The regions in tension must be balanced by those in compression. With D the body's volume:

$$\int {}^{m}\sigma_{ij}\, dD = 0 \,. \qquad (1)$$

This is really nothing more than a statement of mechanical equilibrium.

The second most important point in this discussion of macrostresses is the equilibrium conditions at an infinitesimal element in a body:

$${}^{m}\sigma_{ij,j} = 0, \qquad (2a)$$

$${}^{m}\sigma_{ij}\cdot n_j = 0. \qquad (2b)$$

The indices i, j run from 1-3. Actually, equations 2 can be employed to prove Eq. 1 (Mura, 1982). In all that follows, the axial system in Fig. 4 will be employed. For the moment consider only the specimen axes, P_i and note that the direction "P_3" is normal to the surface of the specimen. One form of Eq. 2a is:

$$\frac{\partial {}^{m}\sigma_{i1}}{\partial P_1} + \frac{\partial {}^{m}\sigma_{i2}}{\partial P_2} + \frac{\partial {}^{m}\sigma_{i3}}{\partial P_3} = 0. \qquad (3)$$

For a macrostress system, ${}^{m}\sigma_{11}$, ${}^{m}\sigma_{12}$ do not vary with P_1 or P_2 and therefore the gradient of ${}^{m}\sigma_{13}$ must be zero. Furthermore, from Eq. 2b, ${}^{m}\sigma_{13}$ is zero at the surface. Hence macrostress ${}^{m}\sigma_{13}$ cannot exist any where in the specimen. In a similar fashion, it can be shown that ${}^{m}\sigma_{23} = {}^{m}\sigma_{33} = 0$.

Finally, the macrostress is, by definition, the same in all phases.

MICROSTRESSES (${}^{\mu}\sigma_{ij}$):

In a processed composite there may be a macrostress, and in the near surface regions, the forces in this region applied by the bulk violate Eq. (2b). Therefore Eq. 3 is no longer true; for microstresses the ${}^{\mu}\sigma_{i3}$ can and do exist. This is the first important concept about the microstresses. The second concept can be obtained by employing Eq. 1 for a two-phase (a and b) material with f the volume fraction of second phase. Integrating over a large volume (Noyan, 1983).

$$(1 - f) \langle {}^{\mu}\sigma_{ij}^{a} \rangle + \langle {}^{\mu}\sigma_{ij}^{b} \rangle = 0. \qquad (4)$$

The carats imply averages over the volume sampled by the integral. This equation has a number of uses. Having measured the microstress in one phase, the microstress can be calculated for the second phase. Secondly, Eq. 4 can serve as a useful check that the measured microstresses are correct (Krawitz, 1984).

These microstresses are balanced locally between phases, between soft and hard regions, between regions with different orientations (and hence different elastic and/or plastic response). When they are present, Eq. 1 may not hold for any one phase. Measurements of stress in one phase vs. depth may even be constant, because the stress is being balanced by another phase. Perhaps this separation of the stress into macro and microstress components is artificial. Yet it helps us to understand why especially in a composite material, the stresses may not appear to balance over a cross section, and why normal components of the stress are sometimes present.

In the next section, we consider how these different kinds of stresses can be measured.

THE EFFECT OF STRESSES ON THE DIFFRACTION PEAK:

The diffraction technique of measuring stresses involves a

measurement of one or more of the interplanar spacings characteristic of a crystalline solid (matrix or particle). This spacing acts as an internal strain guage. The physical principles involved are illustrated in Fig. 5. The interplanar spacing expresses itself in the diffraction pattern through Bragg's law, $\lambda = 2d\sin\theta$, where λ is the wavelength, and 2θ is the scattering angle of the observed peak. Thus, knowing λ, and measuring θ, "d" is obtained. In the example shown in Fig. 5, there is a surface compressive stress, and due to Poisson's effect, the interplanar spacing in the grains diffracting in (a) is larger than in the unstressed state. In (b) the sample has been tilted (ψ degrees). The planes diffraction are now more nearly perpendicular to the stress, and "d" is less than in (a). It is through such measurements that the stress is obtained.

To proceed in a more quantitative fashion, we refer again to Fig. 4. In addition to the sample axes, P_i, we will now be concerned with the laboratory axes, L_i. In particular, measurements are made of the interplanar spacing of the crystal planes perpendicular to L_3, at each Ψ tilt of the specimen. The angle Ψ refers to the tilt of the surface normal relative to the position where this normal bisects the incident and diffracted beams. In the L_i system, strains and stresses will carry a prime, whereas the strains and stresses will be unprimed in the P_i system. Then

with d_o the unstressed interplanar spacing, a straightforward application of tensor algebra to rotate the strains from one axial system to another, and to introduce the stresses leads to:

$$\langle \varepsilon'_{33} \rangle_{\Phi,\psi} = \frac{d_{\Phi,\psi} - d_o}{d_o} = S_2/2\{\langle {}^t\sigma_{11}\rangle \cos^2\Phi + \langle {}^t\sigma_{12}\rangle \sin 2\Phi + \langle {}^t\sigma_{24}\rangle \sin^2\Phi - \langle {}^\mu\sigma_{33}\rangle\}\sin^2\psi + S_2/2\,\langle {}^\mu\sigma_{33}\rangle - S_1\,[\langle {}^t\sigma_{11}\rangle + \langle {}^t\sigma_{22}\rangle + \langle {}^\mu\sigma_{33}\rangle] + S_2/\mathcal{L}_1\{\langle {}^\mu\sigma_{13}\rangle \cos\Phi + \langle {}^\mu\sigma_{23}\rangle \sin\Phi\}\sin|2\psi|. \quad (5a)$$

This equation is for either the matrix or the reinforcement. The superscript t implies a total stress, micro plus macrostress, and the carats imply an average over the volume sampled by the incident radiation. Expressed in terms of strains.

$$\langle \varepsilon'_{33} \rangle = \{\langle {}^t\varepsilon_{11}\rangle \cos^2\Phi + \langle {}^t\varepsilon_{12}\rangle \sin 2\Phi + \langle {}^t\varepsilon_{22}\rangle \sin^2\Phi\}\sin^2\psi + \{\langle {}^\mu\varepsilon_{13}\rangle \cos\Phi + \langle {}^\mu\varepsilon_{23}\rangle \sin\Phi\}\sin|2\psi| + \langle {}^\mu\varepsilon_{33}\rangle \cos^2\psi. \quad (5b)$$

The diffraction elastic constants S_i and $S_2/2$ in Eq. 5a vary with the $hk\ell$ value of the peak and can also vary with processing. Therefore these terms should be measured (Perry, Noyan, Rudnik and Cohen, 1984; Noyan, 1985). Approximate values can be calculated if necessary. [For the (unusual) case of an isotropic material these are $-\nu/E$ and $(1+\nu)/E$ respectively and in this case are independent of $hk\ell$.] It is worth emphasizing that all σ_{i3} components are microstresses whereas all other components are

total (t) stresses, possibly involving both micro and macro-stresses.

The presence of σ_{13} and/or σ_{23} is readily detected because in this case $d_{\Phi,\Psi}$ vs $\sin^2\Psi$ is not linear, Fig. 6, whereas in their absence, linearity is predicted by Eq. 5. If these terms are present, the linear quantities a_1 and a_2 can then be formed from measurements at $+\Psi$ and $-\Psi$ tilts:

$$a_1 \equiv \frac{d_{\Phi\psi+} + d_{\Phi\psi-}}{2d_o} = \frac{S_2}{2}\{\langle{}^t\sigma_{11}\rangle\cos^2\Phi + \langle{}^t\sigma_{12}\rangle\sin 2\Phi + \langle{}^t\sigma_{22}\rangle\sin^2\Phi - \langle{}^\mu\sigma_{33}\rangle\}\sin^2\psi + \frac{S_2}{2}\langle{}^\mu\sigma_{33}\rangle - S_1(\langle{}^t\sigma_{11}\rangle + \langle{}^t\sigma_{22}\rangle + \langle{}^\mu\sigma_{33}\rangle), \quad (6a)$$

$$a_2 \equiv \frac{d_{\Phi\psi+} - d_{\Phi\psi-}}{2d_o} = \frac{S_2}{2}\{\langle{}^\mu\sigma_{13}\rangle\cos\Phi + \langle{}^\mu\sigma_{23}\rangle\sin\Phi\}\sin|2\psi|. \quad (6b)$$

From the linear relation a_2 vs $\sin|2\Psi|$ at $\Phi = 0^\circ$, $\langle{}^\mu\sigma_{13}\rangle$ is obtained and at $\Phi = 90^\circ$, $\langle{}^\mu\sigma_{23}\rangle$.

The term a_1 is obtained at $\Phi = 0^\circ$, 45° and 90°. Calling the slope of a_1 vs. $\sin^2\Psi$, m, then at $\Phi = 0^\circ$, for any phase i:

$${}_om^i = S_2^i/2\ \{\langle{}^t\sigma_{11}\rangle - \langle{}^\mu\sigma_{33}^i\rangle\}d_o^i, \quad (7a)$$

and at $\Phi = 90^{o}$:

$$_{90}m^{i} = S_2^i/2 \{<^{t}\sigma_{22}> - <^{\mu}\sigma_{33}^{i}>\}d_0^i. \qquad (7b)$$

The intercept (I) of the lines for both Φ is invariant and equal to:

$$\frac{I - d_0^i}{d_0^i} = S_2^i/2 <^{\mu}\sigma_{33}^{i}> - S_1^i[<^{t}\sigma_{11}> + <^{t}\sigma_{22}> + <^{\mu}\sigma_{33}^{i}>] \qquad (8)$$

This invariance is a useful check on the data at each Φ.

Equations 7 and 8 can be solved simultaneously to yield $\langle ^{t}\sigma_{11}\rangle$ $\langle ^{t}\sigma_{22}\rangle$ and $\langle ^{\mu}\sigma_{33}\rangle$ for a given phase. However, a value for d_o^i precise in the fifth decimal place is required (Noyan, 1985). This can sometimes be obtained by taking filings and annealing them. Assuming this is possible, then for α and β phases, and keeping in mind that the macroscopic stresses are the same in both phases.

$$<\sigma_{11}^{t}>_{\alpha} = <^{m}\sigma_{11}> + <^{\mu}\sigma_{11}^{\alpha}>, \qquad (9a)$$

$$<^{t}\sigma_{11}>_{\beta} = <^{m}\sigma_{11}> + <^{\mu}\sigma_{11}^{\beta}> \quad , \qquad (9b)$$

and from Eq. 4.

$$(1 - f) \langle {}^{\mu}\sigma_{11}^{\alpha} \rangle + f \langle {}^{\mu}\sigma_{11}^{\beta} \rangle = 0, \qquad (9c)$$

and similarly for $\langle {}^{t}\sigma_{22} \rangle$. Thus $\langle {}^{\mu}\sigma_{11}^{\alpha} \rangle$ $\langle {}^{\mu}\sigma_{11}^{\beta} \rangle$, $\langle {}^{m}\sigma_{11} \rangle$, etc. are obtained and the σ_{22} components at $\Phi = 90^{\circ}$. From a_1 vs. $\sin^2\Psi$ at $\Phi = 45^{\circ}$, data on σ_{12} components are then obtained.

Even if accurate values of d_o are unavailable, it is still possible to obtain useful information. Rewriting Eq. 7a:

$${}_{o}m^{i} = \frac{S_2^i}{2} \{ \langle {}^{m}\sigma_{11} \rangle + [\langle {}^{\mu}\sigma_{11}^{i} \rangle - \langle {}^{\mu}\sigma_{33}^{i} \rangle] \} d_o^i , \qquad (10a)$$

and:

$$(1-f) [\langle {}^{\mu}\sigma_{11}^{\alpha} \rangle - \langle {}^{\mu}\sigma_{33}^{\alpha} \rangle] + f[\langle {}^{\mu}\sigma_{11}^{\beta} \rangle - \langle {}^{\mu}\sigma_{33}^{\beta} \rangle] = 0. \qquad (10b)$$

From Eqns. 10, $\langle {}^{m}\sigma_{11} \rangle$ and the net microstress $\langle {}^{\mu}\sigma_{11}^{i} \rangle - \langle {}^{\mu}\sigma_{33}^{i} \rangle$ can be obtained for each phase (Noyan and Cohen, 1985). With corresponding equations the $\langle \sigma_{22} \rangle$ components are determined.

The approach applied here actually assumed that all the data was obtained with x-ray diffraction with low penetration. Neutrons have a very large penetrating power (several cm in most materials) and with this source, macrostress contributions will

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4) The use of filler to sample the stress state (at various depths) after fabrication and during service is particularly attractive for polymeric matrices.

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did recognize that the sampled stress tensor was three dimensional and employed a shortened version of the techniques described here. The effects of moisture absorption could be detected, as well as the stresses due to fabrication. This technique offers great promise; suitable powders added during forming would permit periodic stress measurements during use. By using different powders at different depths, stress may be sampled vs. depth.

CONCLUDING REMARKS:

1) To date, few measurements of residual stresses in composites have been carried out correctly, taking into account the three dimensional stress tensor, and the decay of stress components toward the surface.

2) A suitable framework for performing such measurements has been described here, that can be carried out in reasonable times. The procedure permits the separation of macro and microstress components.

3) If time does not allow for such details, the most useful measurements are of the lattice parameters of the phases present, from which some idea of the hydrostatic stress component can be obtained.

the net stress was much lower than the predicted value, probably because of this flow. The value reached saturation along the fiber length, which suggested to the authors that such measurements can lead to a measure of the transfer length. Extrapolating values to the very tip they propose could give information on the effective interfacial shear strength, and by comparison with a fiber not completely embedded, information on size effects might be obtained.

Barrett and Predecki have developed a most interesting technique for this class of materials (Barrett and Predecki, 1976, 1978, 1980; Predecki and Barrett, 1979, 1982) useful even if neither the matrix or reinforcement yield suitable diffraction peaks, such as graphite or glass fiber reinforced epoxy resins. Small amounts of a crystalline powder are dusted on the ply. By choos-sing different powders, the stresses at various depths may be sampled. The criteria for choosing a powder, besides having a suitable peak at high angles, is that this material have a yield strength higher than the maximum expected in the composite. Smaller amounts must be used than were employed in these studies to avoid large effects on the local stress pattern. Reductions by a factor of ten should be possible by employing PSD's While the authors could not separate macro and microcomponents with diffraction from only one phase in a three-phase solid, they

tension, as expected, as the carbide contracts much less than the Aℓ alloy matrix.

Both Ledbetter and Austin (1986) and Chipman (1975) recognized the three dimensional nature of the stress system, but not the fading near the surface, nor the difference between micro and macrostresses.

It will be interesting to explore many of these results with the more quantitative procedures described here.

REINFORCED AMORPHOUS MATRICES:

The studies on glass with embedded thoria and alumina (Grossman and Fulrath, 1961) have already been discussed. Of particular interest in this section are studies of reinforced resin matrices.

Gillin et. al. (1969) and Hawkes (1974) measured the stresses along a Pt-Rh wire, 10 mm long, embedded in a casting resin cured at 323° K. Net stresses along the wire were measured with the wire totally embedded in the resin, and also with one tip free. Comparison with theory indicated that there was plastic flow and

in Fig. 1. These authors also reported that the plastic flow accompanying a quench from room temperature into liquid nitrogen reduced the net stress, perhaps by increasing the hydrostatic component.

Chipman (1975) measured the difference in net stress in the matrix parallel and transverse to the $A\ell_2O_3$ fibers in Aℓ and Mg matrices. This value was large and positive. Chipman employed transmission rather than the usual reflection techniques. Scatter due to grain size was troublesome. (Small oscillations can help with this problem.) As the author suggests, the use of sychrotron radiation could be especially useful. A wavelength can be chosen to optimize penetration. Transmission has one useful feature. While it is always best to use a peak at high 2θ to maximize the peak shift for a given strain, it is sometimes difficult to find suitably intense peaks in this region, especially from the reinforcement, and a peak at modest angles must be chosen. In reflection geometry, instrumental errors increase with decreasing θ, but these decrease in transmission.

Ledbetter and Austin (1986) studied both the matrix and particles of 6061 Aℓ reinforced with particulate SiC. They examined the value of d at $\psi = 0$, in effect measuring $\langle {}^{\mu}\epsilon^{i}_{33} \rangle$ in both phases. The SiC was in compression while the Aℓ was in

METAL MATRIX COMPOSITES:

Samoilov et. al. (1977) examined Aℓ alloys reinforced with boron fibers. Only the stresses in the matrix were sampled (the net stresses from slopes of ϵ' vs $\sin^2\psi$) as a function of extrusion temperature and time. The net stresses were of the order of the yield strength of the matrix and reversed in sign (from tensile to compressive) when samples were quenched from room temperature to liquid nitrogen. This does not really imply a change in the sign of the stress, but rather a change in the net stress such that σ_{33} was larger or smaller than σ_{11} (or σ_{22}).

Tsai et. al. (1981) and Tsai (1980) examined Aℓ alloys reinforced with variously coated graphite fibers. These were inserted in the melt of the matrix and the authors pointed out that the difference in contraction along the length of the fiber and the matrix was sufficient to cause plastic upset. Therefore although there was little difference in the contraction transverse to the fibers, significant net stresses were also observed in this direction due to the plastic flow. A particular feature of this investigation was the attempt to sample the interface region close to the fiber, and the middle of the region between fibers, by electropolishing. Much lower net stresses were found in the latter region, confirming that a strong gradient exists as

measured volume (as with neutrons) might be a better way to examine stress vs. depth.

We turn now to results for specific composite systems. Before doing this however, it is necessary to reiterate that except for the α/β brass already discussed, no complete measurements are available. Most investigators have examined only the slope of $\langle \epsilon'_{33} \rangle$ vs. $\sin^2\psi$, which yields a net stress, a net which includes micro and macro components as well as the difference in stress in different directions. It is not surprising that in such a case the value may be much smaller than theoretical expectation - and even the wrong sign. Nevertheless, many results provide at least a useful qualitative picture and the net stress may be what is desired. If time is of the essence, the best measurement to make is the peak position at $\psi = 0$ - which, as already stated, yields $\langle {}^{H}\epsilon^{i}_{33} \rangle$ (Eq. 5b). If d_o is known this takes only seconds. Such a measurement will not provide quantitative information on the near surface stress state, but if it is believed that in the bulk the stresses are hydrostatic this measurement provides at least the correct sight of the strain value of interest, if not the correct magnitude (due to strain relaxation in the near surface region). And if the correct wavelength is chosen the correct value of $\langle {}^{H}\epsilon^{i}_{33} \rangle$ can be obtained.

absorption of the beam by the particles and matrix, and hence on the particle size, spacing and choice of radiation. Positive or negative slopes will occur if only the region of decay is sampled. This is illustrated in Fig. 9. Thus, in making diffraction measurements, it is necessary to decide if stresses near the surface are most important, or if the bulk values are of more interest, and to choose a wavelength appropriately, as well as the analysis technique.

Grossman and Fulrath (1961) is particularly interesting because the coefficient of expansion of the glass matrix could be adjusted to be larger or smaller than the suspended alumina or thoria. With alumina, horizontal lines of ϵ' were observed for the alumina, with a positive intercept when the expansion of the glass was less than the alumina and vice versa.

Because the normal components decay as the surface is approached and because these components can be very important in composites, caution is required in surface preparation for stress measurements. Polishing will produce stresses in these regions. Upon etching, the normal stresses adjust to zero at the new surface, affecting all the stress components. If depth is to be probed the use of short wavelengths and slits to define the

ring time. While the peak position has been found by parabolic fitting to the uppermost regions of a peak (James and Cohen, 1977), the entire peak is recorded in a PSD. It has just been shown that fits to the entire peak with a modified Lorentzian function can provide the same precision as a parabolic fit, but in a much shorter time (Devine and Cohen, 1986). This procedure has the added advantage that only slightly more than half the peak is required. With composites, peaks from the matrix and support could overlap, an this feature could prove helpful.

When the origin of the stresses is differential thermal contraction during forming, the stresses may be nearly hydrostatic. In such a case the various plots (a_1 or ϵ'_{33}) should be horizontal with intercepts that yield the values of strain component $\langle {}^{\mu}\epsilon_{33} \rangle$ which by itself is adequate to describe a hydrostatic stress state. However, the microscopic normal components are zero exactly at the surface and hence decay to this value as the surface is approached. This has been modeled theoretically and experimentally (Nishioka, Hanabusa and Fujiwara, 1974; Takei, Nishioka, Hanabusa and Fujiwara, 1977; Hanabusa, Nishioka and Fujiwara, 1983; Grossman and Fulrath, 1961; Krawitz, 1985) (Si in Al, carbides in steel, alumina and thoria in glass, WC in Co). Krawitz, (1985) shows that whether or not sufficient depth is sampled to see the full hydrostatic state depends on the

vs. $\sin^2\psi$ (if closely spaced ψ values are used) for some reflections but not others. The fluctuations can be employed to estimate this coherency strain.

VARIOUS TECHNIQUES:

To follow the approach described here data must be obtained at $\phi = 0^\circ$, 45° and 90° and for each ϕ at, say, six or more ψ tilts. With normal x-ray detectors this can take about two-three hours for a specimen. With neutrons several days can be involved. However, there is a new class of detectors called position sensitive detectors (PSD) that record the entire peak at one time - a kind of digitized film. These are available for x-rays (James and Cohen, 1976; James and Cohen, 1984), and neutrons (Tompson, Mildner, Mehregany, Sudol, Berliner and Yelon, 1984). These reduce the measuring time with x-rays to the order of one hour or less, (and with neutrons to the order of a day). There are now several examples of the determination of the entire 3-D stress tensor, Dölle and Cohen (1980) for example, and determination of macro and micro components in two phases has been demonstrated in brass (Noyan and Cohen, 1985). The results from this last reference are illustrated in Fig. 8.

Another step has recently been taken which reduces the measu-

Occasionally, anomalies have been reported in this type of data. For samples with strong preferred orientation, large oscillations sometimes appear in $\langle \epsilon_{33} \rangle$ or a_1 vs $\sin^2\psi$. (This is why data at several ψ is required, to check for such anomalies.) These are of two origins: 1) They can be due to elastic anisotropy, in which case switching to an 00h or hhh reflection may help (Dölle and Cohen, 1980). 2) The origin may be in the fact that there are variations in microstress from point to point; perhaps some regions yield more easily than others. No complete analysis is possible in such a case. The average values are not of much use in the presence of large fluctuations (Noyan and Cohen, 1984). Two methods exist in such a case for obtaining some information on the range of stresses present; 1) either from the data on peak position itself (Marion and Cohen, 1975) or 2) from the Fourier analysis of the shape of the x-ray peaks (Schwartz and Cohen, 1977); with this latter procedure the variance of microstrain is obtained. It has been shown that the use of shorter wavelengths may eliminate these oscillations (Hauk and Vaessen, 1984) but this just means that the stresses themselves are averaged over too large a depth.

Doig and Flewitt (1985) have just shown that if the second phase is aligned with some preferred orientation and there are appreciable coherency strains, large fluctuations will occur in d

average to zero and only the microstresses affect the results. The entire microstress tensor can then be obtained with a_1 and a_2 at three Φ's. By restricting the beam with slits, Fig. 7, macrostresses are again included and with the total terms and the known microstresses, the macrostresses can be obtained. However, this can only be done in the absence of strong preferred orientation (Noyan and Cohen, 1984).

If ϵ_{13}, σ_{23} are absent, all the above procedures are simplified as the analysis can proceed (in the same way) with $\langle\epsilon_{33}'\rangle$ at three Φ's; the terms a_1 and a_2 are not required.

There are many useful aspects of employing diffraction. Firstly it is nondestructive. Secondly portable units exist for studies in the field (James and Cohen, 1980). Thirdly, errors have been evaluated for instrumental effects (James and Cohen, 1980) for counting statistics (James and Cohen, 1977; Rudnik and Cohen, 1986) and for the elastic constants (Perry, Noyan, Rudnik and Cohen, 1984). Therefore it is not only possible to evaluate the errors in a measurement without repetition, but it is possible to automate the measurement to an operator specified precision (James and Cohen, 1977).

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FIGURE CAPTIONS

FIGURE 1: Schematic of the residual stress pattern in the matrix between bonded fibers. The dotted line is the average sampled in most diffraction measurements.

FIGURE 2: Top: The longer slab is compressed and joined to the shorter one. When the stress is removed, the two slabs stress each other, the top one being put into compression by the bottom one, and vice versa.

FIGURE 3: A typical macrostress gradient as a function of distance from the surface.

FIGURE 4: The axial system employed in this work. The p_i describe the sample, whereas the L_i describe measuring space. The interplanar spacing of planes normal to L_3 are measured in a stress determination.

FIGURE 5: Schematic of a diffractometer for stress measurement. (a) Certain atomic planes satisfy Bragg's law and diffract x-rays at a 2θ value which depends on the spacing of the $hk\ell$ planes. This spacing is affected by stresses. (b) After the specimen is tilted, diffraction occurs from other

grains but from the same set of planes. Since the normal stress component on these is different than in (a), the plane spacing will be different as will the diffraction angle.

FIGURE 6: Interplanar spacing of the 211 peak (taken with CrK_{α} radiation) vs sin2θ for a ground steel. From Dölle and Cohen (1980).

FIGURE 7: The volume sampled by a neutron beam is defined by slits. By moving the slits or the specimen, different regions can be measured.

FIGURE 8: Separated macro (m) and micro (p.m.) stress components vs depth. Shot peened α/β brass; $\theta = 0$. The $\sigma^{p.m.} = (\langle {}^{\mu}\sigma_{11}\rangle - \langle {}^{\mu}\sigma_{33}\rangle)$. From Noyan and Cohen (1985).

FIGURE 9: Calculated interplanar spacing for the 122 WC peak for CuK_{α} radiation. WC with 10 pct Co. Interparticle spacing of 1, 2, 4 and 10 μm were employed with a hydrostatic compression of 400 MPa. (Krawitz, 1985).

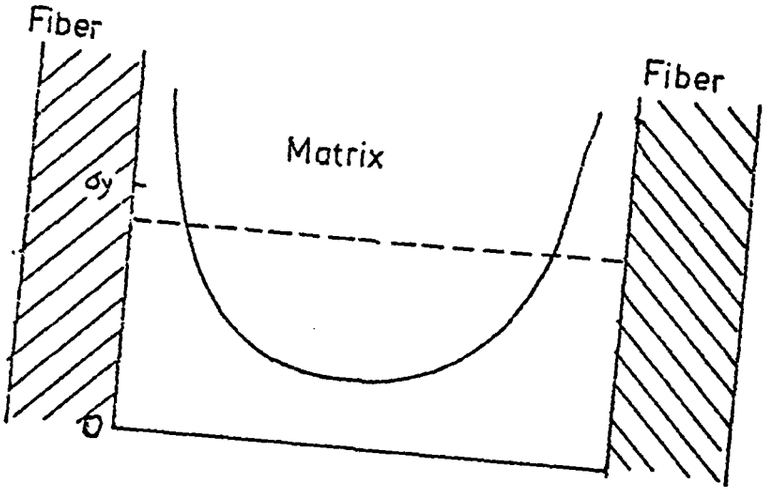


Figure 1: J. B. Cohen

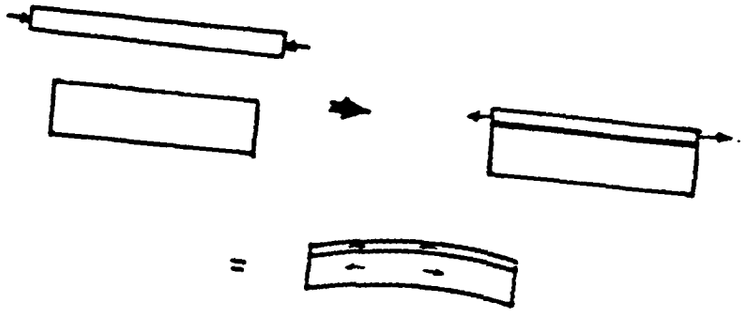


Figure 2: J. B. Cohen

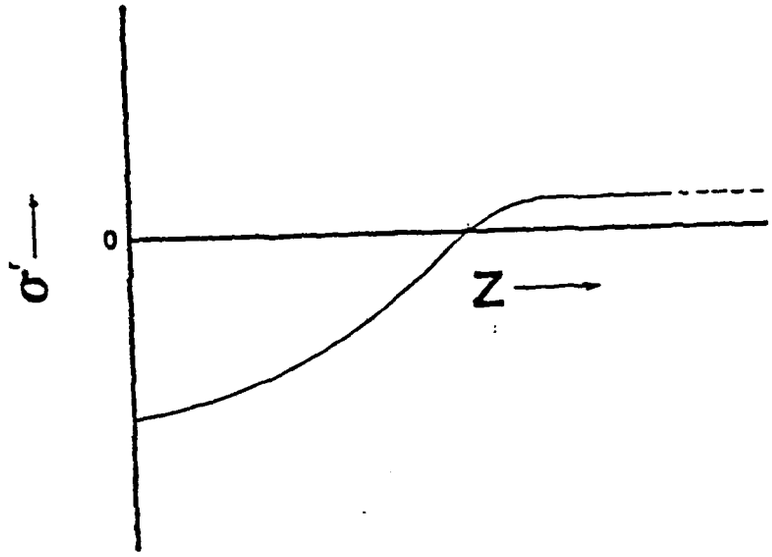
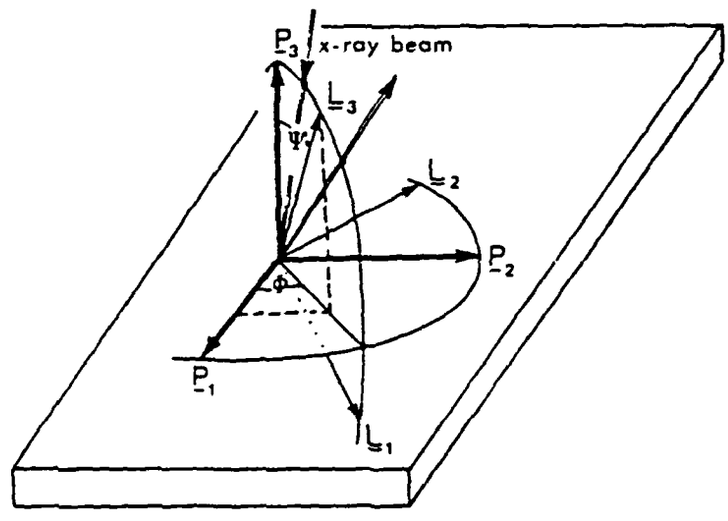


Figure 3: J. B. Cohen



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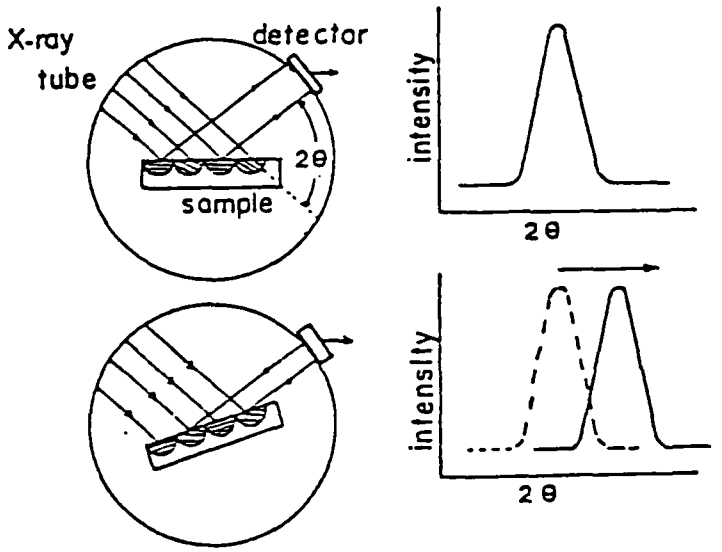

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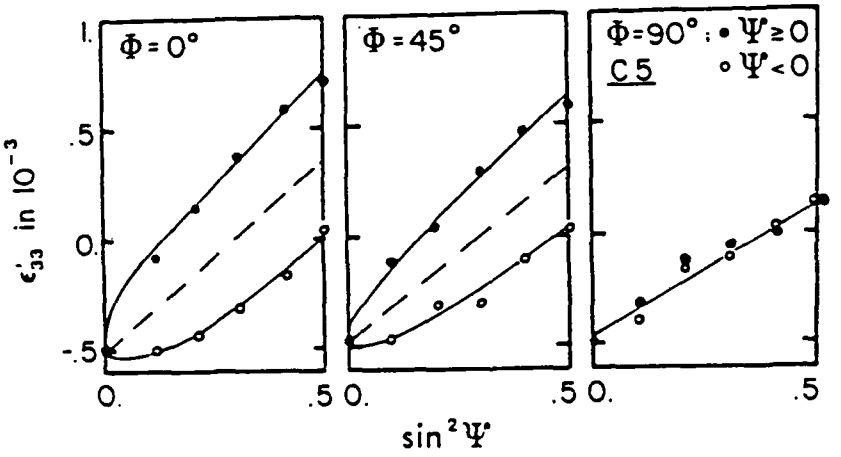
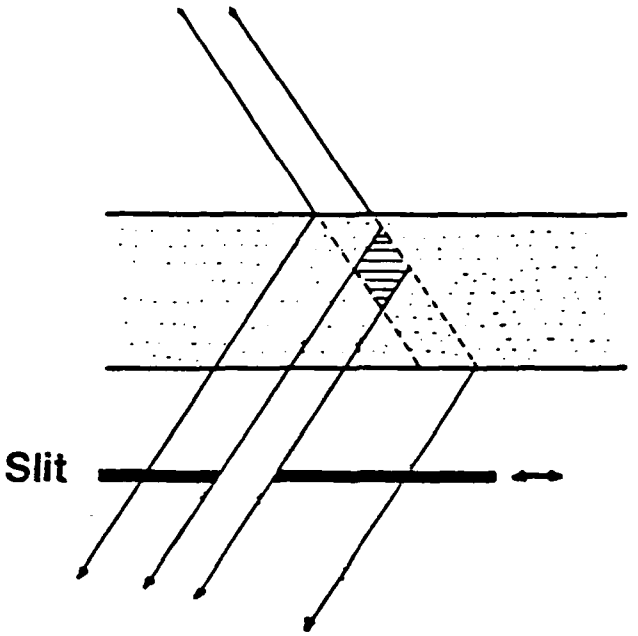


Figure 6: J. B. Cohen



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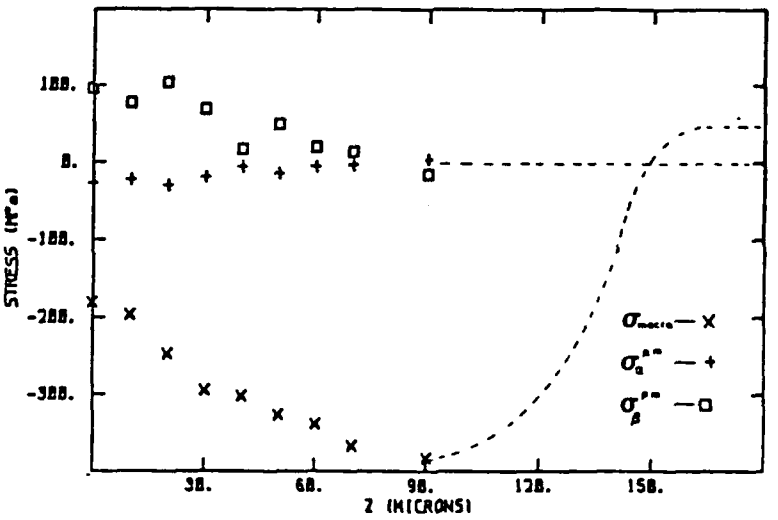
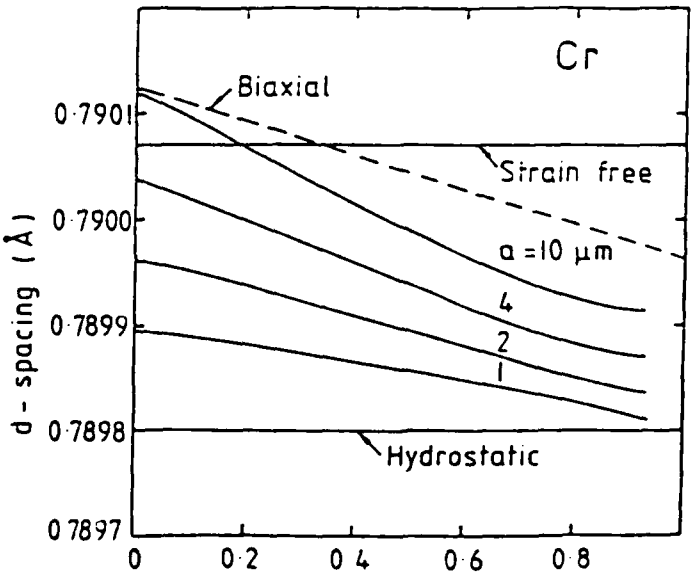


Figure 8: J. B. Cohen



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ABSTRACT

Although there is mounting interest in the measurement of stresses in composite materials after fabrication and/or use, few measurements to date have not taken into account the three dimensional nature of the stress system in such materials. Most data gives only the net stress, that is the difference between principal stresses. A procedure for a more complete measurement (in a reasonable time) is developed here, including the separation of macrostresses and microstresses. If times does not permit a full investigation, measurements of the lattice parameters of the component phases provide a simple way to sample the hydrostatic component due to differential thermal contraction. The Barrett-Predecki method of adding filler is particularly promising for stress measurements in those composites whose component phases do not give appropriate diffraction peaks. This procedure could also be used for monitoring streses during the useful life of such materials.

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